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SIGNIFICANCE OF PLASTICITY CHARACTERISTICS REFLECTING
CERTAIN ASPECTS OF THE PHYSICAL STATE OF ALLOYS

Zhurnal Tekhnicheskoy Fiziki
[Journal of Technical Physics],
Vol XXIV, No 3, 1954,
Pages 425-432

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The maximum plasticity of most alloys, as determined during elongation, consists, as is well known, of 2 substantially different components: (a) uniform deformation which develops up to the moment when the stretching force attains its maximum value and is characterized by a relatively uniform distribution along the length of the specimen being stretched and (b) concentrated deformation which becomes localized after the stretching force passes through a maximum on a comparatively small section of the length of the specimen.

It was shown in the work (1) that the value of the maximum deformation during stretch depends substantially on the state and characteristics of the metal and determines its capacity for intensive physical strengthening and also characterizes the extent of cold hardening. On the basis of this, the conclusion was arrived at regarding the expediency of measuring the value of the maximum uniform deformation during mechanical tests.

The indicated results of the work (1) were utilized successfully in the analysis of the mechanical state of small crystals of α -iron of high-tempered martensite (2). Further investigations have shown that not only the uniform but also the concentrated component of the deformation reflects certain aspects of the physical state of the alloys.

Below are given experimental data which prove this and which indicate the expediency of individual examination of the uniform and concentrated deformation.

Let us examine the influence of carbon on the plasticity of steel.

Figure 1 shows curves which illustrate the nature of the change in the uniform and concentrated deformation as a function of the carbon content in the steel subjected to quenching and subsequent 1.5-hour tempering at 600° C. The quenched and high-tempered carbon steel represents a 2-phase system -- a conglomerate of small crystals of α iron and small crystals of cementite distributed in the iron. It can be assumed that for the same tempering temperature, the α iron in steels of different carbon content is in the same mechanical state. This is confirmed by the data of Kurdjumov and Oslen (3) which indicate a constant value of the strains of second order in steels of different carbon content which were tempered after quenching at the same temperature. Thus an increase in the carbon content in the test specimens (Figure 1) can lead only to an increase in the amount of the carbide phase in the conglomerate α -Fe-Fe₃C.

Returning to Figure 1, one can become convinced that an increase in the amount of carbides in the steel does not influence the value of the uniform deformation but decreases substantially the concentrated component of the deformation, which is determined during stretch.

A similar relationship between the behavior of the uniform and concentrated deformation takes place also at other temperatures of tempering within the interval of 400-680° C.

Figure 2 shows the curve of the variation in the yield point, σ_y , of the relatively concentrated narrowing zone, and of the true resistance to ductile rupture S_k with increasing carbon content in steels quenched and tempered at 600° C. As can be seen from the diagram, the resistance of the steels to plastic deformation increases with increasing carbon content up to 0.45%, after which it decreases somewhat. Such a nature of change in the yield point is directly related with the quantitatively measurable characteristics of the carbide phase (4). It follows from Figure 2 that for different carbon contents in steels treated under the same conditions it is possible to obtain the same yield point.

By taking advantage of this peculiarity of the influence of carbon, we selected specimens (with 0.38% C, tempered after quenching at 670° for 15 hours and specimens with 0.81% C, tempered also at 670°, but for 3 hours only) in such a manner that the deformation stresses during the testing of medium and high-carbon steels practically coincided. The difference between the tested steels was expressed in the amount of disperse carbides contained therein. The difference in the behavior of the specimens during stretch was expressed by the fact that the deformation curve of the specimens containing a greater amount of disperse carbides broke off earlier due to untimely destruction (Figure 3). In order to determine the essence of the influence of the disperse carbides on the plasticity of the steel, we shall return to Figure 4.

Figure 4 shows the deformation curves obtained by Pavlov and Yakutovich by twisting specimens of 60S2 steel which were quenched and tempered at 500° C and which had and did not have microscopic cracks after quenching (5). As follows from Figure 4, the curves

of the plastic flow of the 2 types of specimens coincide entirely. The influence of the microcracks is evidenced only by the fact that the deformation curve of the specimens with microcracks breaks off earlier due to the untimely destruction by cutting. At tempering temperatures which give ductile destruction the existence of defects in the crystal lattice in the form of microscopic cracks was evidenced only in a decrease of the maximum deformation which was attained at the moment of destruction.

By comparing Figures 3 and 4 it can be seen that the influence of the carbides is similar to the influence of defects in the crystal lattice which existed in the steel prior to the start of the testing. The dispersed carbides included in the α phase favor apparently the development of defects in the latter which lead to the ductile destruction at more early stages of the plastic flow. The mechanism of such action of the carbides is as yet not clear. It is possible that such a mechanism is related to the redoubling in the uniformity of the plastic flow. The results of the microscopic investigation of the destruction zone of ruptured specimens are in accord with this assumption. The destruction, as a rule, occurs and proceeds along the ferrite, rounding the carbide; only in rare cases does it pass along the boundary of division between the phases.

It follows from Figure 2 that the relative narrowing with increasing carbon concentration in the steels changes regardless of the change in other characteristics, including also the resistance to plastic deformation. The data shown in Figures 5, 6, and 7 also serve as confirmation of this.

Figure 5 shows a curve of the yield point of the 2-phase system α -Fe+Fe₃C as a function of the parameter $V_f = \frac{1-(C/6.68)}{1.5}$,
(MD)
where M and D are the amount of carbides per unit volume and their diameter, respectively (4).

The attempt to establish a correlation between the concentrated narrowing and the same parameter V_f ended in complete failure; this is indicated by Figure 6.

A satisfactory correlation was obtained between the concentrated narrowing and the other parameter which is a function of the diameter and the total surface of the carbides (Figure 7),

$$P = sD^{2/3}.$$

This difference indicates the lack of a relationship between the strengthening action of the carbides and the mechanism of the influence of the carbides on the plasticity of the steel.

Figure 1 shows data which indicate no influence of the carbides on the value of the uniform deformation of the tempered steel.

By increasing the tempering temperature within the interval of 400-680° C, the uniform deformation increases sharply. The change in uniform deformation during tempering is determined by the characteristics of the α phase and characterizes the extent of cold hardenability of the small crystals of α iron (2).

As is well known (6), during quenching, alloy iron undergoes strengthening -- cold hardening of the small crystals of α iron due to the deformation of the crystal lattice of the alloy as a result of the spatial changes during the polymorphic transformation $\gamma \rightarrow \alpha$.

During the process of tempering quenched specimens, the alloy loses strength -- the resistance to plastic deformation drops and the plasticity rises.

Figure 8 shows the curves of the change in the true characteristics of the uniform and concentrated deformation with an increase in the duration of tempering at 680° C of quenched specimens of the alloy Fe+1% Cr+4% Ni+0.02% C.

As can be seen from Figure 8, the value of the true concentrated deformation increases as the duration of tempering the quenched alloy increases up to one hour. With a further increase in the duration of tempering up to 15 hours, the value of the concentrated deformation decreases. For this same time interval, from 0 to 15 hours, the value of the true uniform deformation increases continually.

The increase in the uniform deformation during the tempering process is related, apparently, with the liquidation of the structural aftereffects of the cold hardening. The drop in ψ_{conc} during prolonged tempering can be the result of only the growth of the small crystals of the α iron.

The tempering of quenched specimens exerts a similar action on the change of resistance to rupture. Figure 9 shows a curve of the change in resistance to rupture δ_{rup} of the alloy Fe+8% Cr+0.02% C as a function of the tempering temperature of quenched specimens. (The notched specimens with a diameter at the notch of 1.4 mm were tested by stretching in a medium of liquid nitrogen by using the Shevener micromachine with optical recording of the stretch diagram.)

It follows from Figure 9 that the resistance of the alloy to rupture increases as the tempering temperature increases to 300-400° C and drops during tempering at temperatures above 500° C.

By comparing all possible processes taking place in quenched specimens of alloy iron during tempering within the interval of 500-675° C, the drop in the resistance of the alloy to rupture as the tempering temperature increases above 500° C can be accounted for by an increase in the dimensions of the small crystals of the iron. A similar regularity in the change of the resistance to rupture during the tempering of carbon steels was obtained by Bokshteyn (7).

Thus the study of the change in the value of the concentrated deformation (Figure 8) makes it possible to obtain additional information about the processes which take place in alloys during heat treatment. On the curve showing the complete change in the plasticity of the alloy, the uniform and concentrated deformation, which change in opposite directions, level off the action of each other.

It was established in works (8, 9) that considerable internal strains occur during the quenching of alloy iron and steel. It can be assumed that, at the moment of completion of the phase transformation, the system will not be in a state of equilibrium. Under the influence of the strains which had just been formed but which did not yet become balanced -- which in individual microvolumes exceed the elastic limit of the products of transformation -- there will take place a rearrangement of the lattice which will lead to resorption, in the first place, of the local strains which are distributed, most likely, along the boundaries of the grains.

The mechanical characteristics of these steels, as is well known, have a high sensitivity to structural changes in alloys.

Figure 10 shows the curves of the change in the basic mechanical characteristics of alloy iron containing 8% Cr and 0.08% C, obtained during stretch, as a function of the period of holding the quenched specimens at room temperature (10).

It follows from the diagram on Figure 10 that the value of the yield point and of the uniform deformation does not depend on the length of the holding period after quenching. The values of the concentrated narrowing and of the true resistance to rupture S_k increase strongly. The investigation of the problem has led to the conclusion (10) that the most probable cause for the sharp decrease in ϵ_{conc} and S_k as a result of the quenching and of their growth during subsequent relaxation are to be considered the non-uniform internal strains which attain the maximum value on the ends of the grains. The considerable local strains can be regarded as defects of the crystalline lattice, from which original cracks develop during the process of plastic deformation and which lead to the untimely destruction of the specimens.

The relaxation of nonuniform local strains which take place at room temperature with time leads to the liquidation of the indicated defects and, as a result, to the restoration of the normal value of ϵ_{conc} and S_k , without, at the same time, influencing the extent of strengthening.

Experiments which were carried out with carbon-free iron alloy were repeated with carbon steel containing 0.17% C. Sharply quenched specimens of carbon steel suffered brittle destruction during stretch.

The resistance of the specimens to rupture increased sharply prior to testing with an increase in the holding period after quenching, as this follows from Figure 11. Thus the data on Figure 11 also confirm the view about the occurrence of defects in the lattice during quenching and their liquidation with time by the relaxation processes. The indicated defects decrease the value of the concentrated narrowing of the alloy during stretch and do not exert any influence on the value of the uniform distribution and of the resistance to plastic deformation.

Consideration of the results

The following peculiarities in the behavior of the characteristics of uniform and concentrated deformations during the variation of different factors related with the composition and physical state of the alloys deserve attention.

1. The uniform and concentrated deformations vary frequently, independently of each other; this indicates a different physical content of these characteristics (Figures 1, 8, and 10).

For example the rearrangement of the lattice, which causes a drop in ψ_{conc} during quenching of the alloy iron and the subsequent growth of ψ_{conc} with time is directly not related with the rearrangement of the lattice, which causes the strengthening (cold hardening) (Figure 10). This is indicated by the lack of a relationship between the strengthening action of the carbides and the mechanism of the influence of the carbides on the plasticity of the steel (Figures 5-7).

2. The uniform deformation is a characteristic sensitive to cold hardening; the relative drop in the capacity of the metal for uniform deformation can be regarded as a measure which characterizes the relative value of the physical strengthening acquired by the metal.

When the strengthening is attained by cold hardening, the value of the uniform deformation varies in accordance with the change in the yield point (1, 2, 6).

3. The concentrated deformation is a characteristic sensitive to defects in the crystalline lattice. The occurrence of defects leads to destruction at much earlier stages of the plastic flow; the value of the concentrated deformation thereby decreases (Figures 3 and 10).

Numerous observations show that the presence of metallurgical flaws in the steel, for example, flakiness in high-alloy steels, has no influence on the uniform deformation but noticeably decreases the value of the concentrated deformation.

On the basis of the available experimental data it can be assumed that the completion of the plastic deformation with ductile destruction during stretch is the result of not so much of an increase of the active strains as the result of the development of defects in the crystal lattice during the process of the very plastic deformation which takes place under definite conditions of strained state.

The development of the defects at a well-known stage of the plastic deformation leads to the formation of an original crack which leads to destruction.

Such a picture is in agreement with experimental data. It is well known that the plastic deformation leads to the formation of submicroscopic defects in the crystal lattice (11), loosening of the lattice, which is evidenced by a gradual drop in the density of the alloy (12), and the formation of micropores and microcracks. The works by Rebinder and his associates have shown that, on the basis of the defects during the process of plastic deformation, there develop microscopic slits and cracks (13). Having observed microscopically the mechanism of destruction during plastic twisting, Yakutovich and Rybalko (14) have discovered that, starting with a certain moment, the deformation becomes localized in a narrow zone in which the metal becomes loose; microvoids are formed which change into macrovoids and then blend into a common crack which leads to fracture.

Fridman and Zilova observed the formation of cracks during twisting, at much earlier stages prior to the localization of the deformation (15).

Conditions exist which favor the development of the indicated defects during the process of plastic deformation and which lead, as a result of this, to a much earlier destruction. Such conditions can be created for example by a considerable nonuniformity of the plastic flow, which is most probable in heterophase alloys; by a change in the strained state, which is determined by the relationship of the acting normal and tangential strains; by the presence of local distortions of the lattice; and finally, by the presence of dangerous defects even prior to the start of the plastic deformation. A classic example of such an influence of obvious defects present in the metal prior to the deformation is Figure 4.

The factors which cause the ultimate destruction in many cases exert influence only on the change in the value of the concentrated flow and do not exert influence on the value of the uniform deformation. This makes it possible to assume that, on the section of the concentrated deformation, there develops primarily a rearrangement of the atoms which leads to the development of defects in the crystal lattice.

And thus, if the chemical composition of the alloy, its structure, or the conditions of heat treatment are varied in a regular manner, then the change in the uniform deformation will, as a first approximation, indicate that cold hardening of the crystal lattice has taken place; however, a decrease in the value of the concentrated deformation will indicate either the occurrence of defects in the crystal lattice or the formation of conditions which favor their more early development during the process of plastic deformation.

From all that was said above it follows that during systematic investigations of the influence of different factors on the physical state of alloys it is expedient to make a individual examination of the uniform and concentrated components of the deformation.

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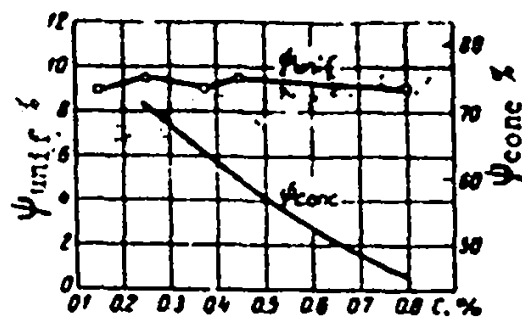


Figure 1. ψ_{unif} , uniform narrowing; ψ_{conc} , concentrated narrowing; C, carbon content

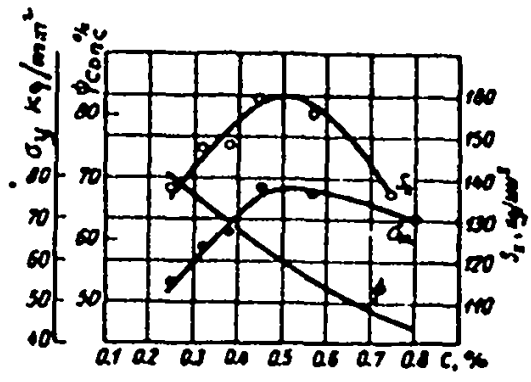


Figure 2

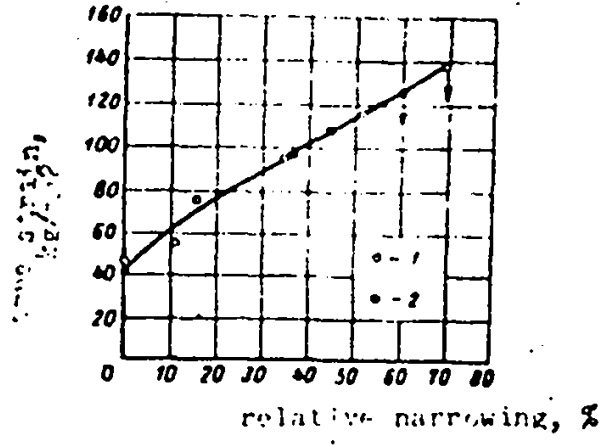


Figure 3. Curves of deformation by stretching specimens of 2 carbon steels with a carbon content equal to 0.3% (1, steel 1) after quenching and tempering at 200°C for 15 hours and with 0.1% C (2, steel 2) after quenching and tempering at 200°C for 3 hours

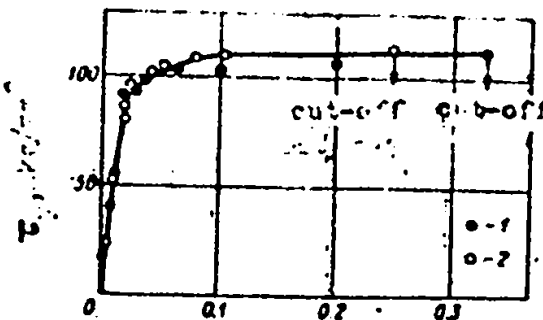


Figure 4. Curves of deformation by twisting specimens of 60C2 steel, quenched and tempered at 500°C and which did not have micropores after quenching

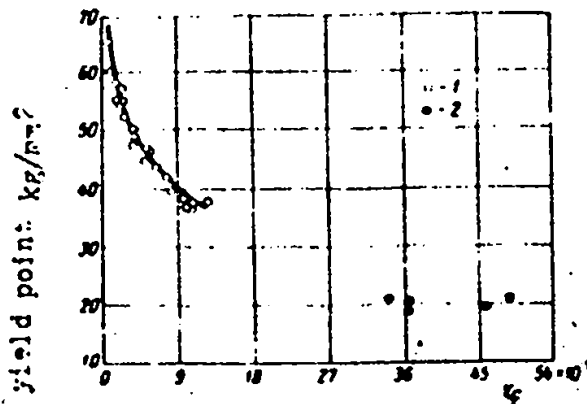


Figure 5. Relationship between yield and effective volume of V_f for steels containing globular carbides; 1, sorbite and globular perlite after quenching and tempering; 2, spheroidized perlite

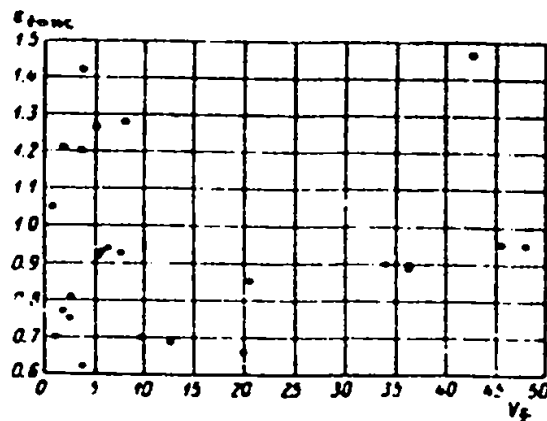


Figure 6. Relationship between true concentrated deformation (true elongation $\epsilon_{conc} = \ln \frac{1}{1 - \psi_{conc}}$) of carbon steels with different carbon content and parameter V_f . (Specimens tested were quenched and tempered (at 500°C) for different periods of time)

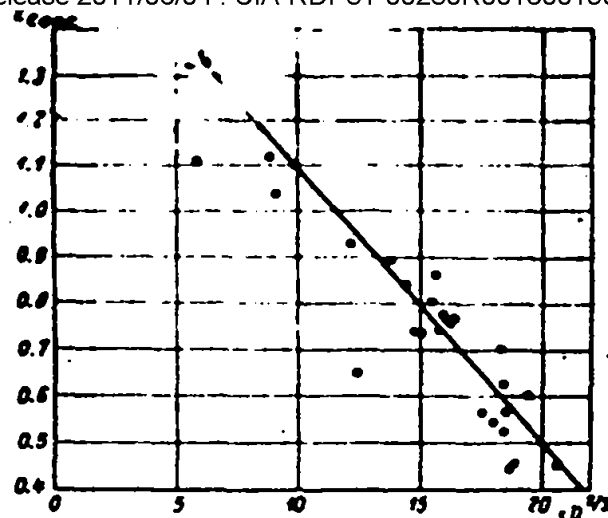


Figure 7, Relationship between value of true concentrated deformation (true elongation $\epsilon_{conc} = \ln \frac{1}{1 - \psi_{conc}}$) of carbon steels with different carbon content and parameter $R = 3D^2/3$, where s is the surface of the carbide phase and D is the average diameter of the carbides. (The state of the test specimens is the same as on Figure 6)

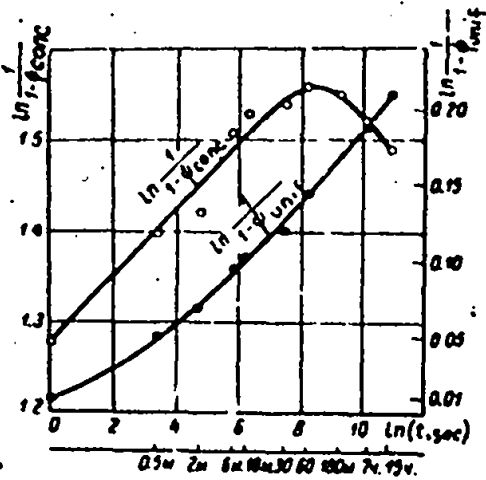


Figure 8

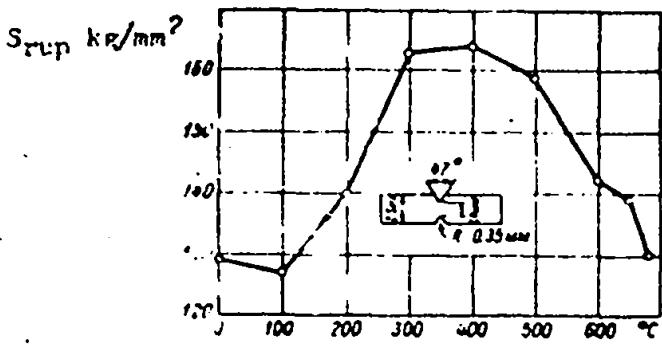


Figure 9

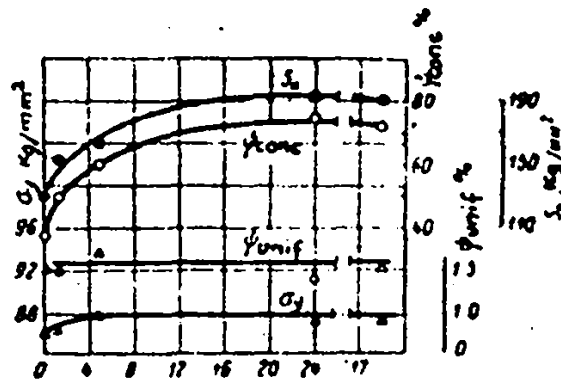


Figure 10. Dependence of characteristics (yield point σ_f [sic], true resistance to rupture S_k , uniform ψ_{unif} and concentrated ψ_{conc} narrowing of quenched specimens of ferrochrome) on the duration of holding at room temperature from the moment of the quenching of the specimens to the start of their testing by twisting. Each point of the diagram is the average of 4-6 tests

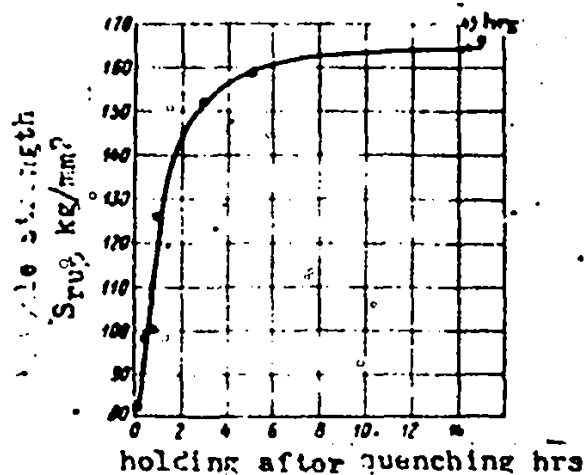


Figure 11. Dependence of resistance to rupture of quenched specimens of carbon steel with 0.17% C on the duration of holding at room temperature from the moment of the quenching of the specimens to the start of their testing by twisting. Each point of the diagram represents the average result of 2, to 4 tests